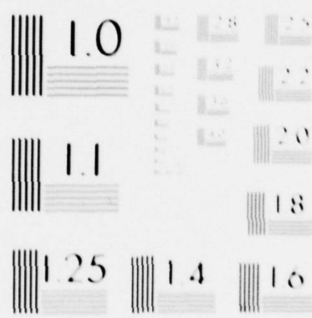
AD-A070 296 DAVID W TAYLOR NAVAL SHIP RESEARCH AND DEVELOPMENT CE--ETC F/G 11/6
FATIGUE FAILURE PREDICTION BY X-RAY DOUBLE CRYSTAL DIFFRACTOMET--ETC(U)
JUN 79 R N PANGBORN, S WEISSMANN, I R KRAMER

UNCLASSIFIED DTNSRDC-79/063 NL

| OF |

AD
A070296

END
DATE
FILMED
7-79
DDC



MICROCOPY RESOLUTION TEST CHART

LEVEL

12

DAVID W. TAYLOR NAVAL SHIP RESEARCH AND DEVELOPMENT CENTER

Bethesda, Md. 20084

FATIGUE FAILURE PREDICTION BY X-RAY DOUBLE CRYSTAL DIFFRACTOMETRY

by

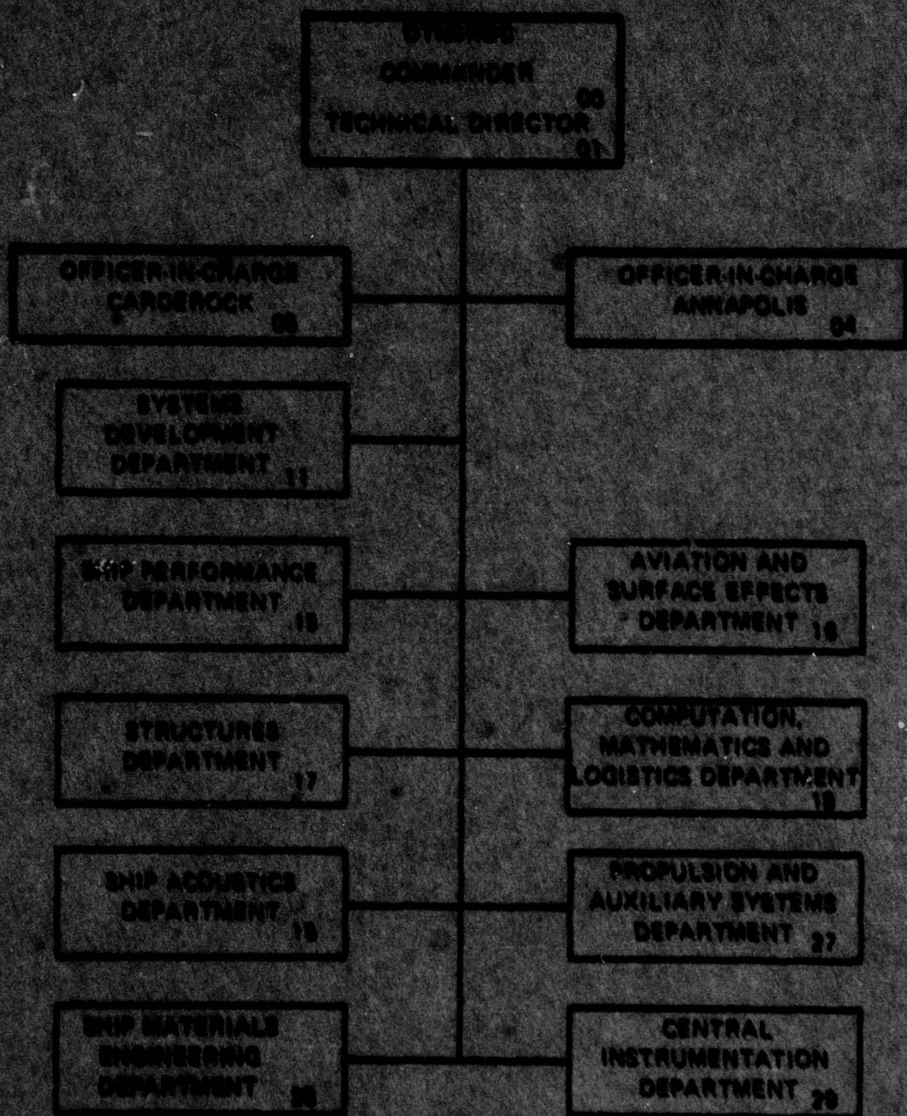
R.N. Pangborn and S. Weissmann
Rutgers University
Piscataway, NJ 08854
and
I.R. Kramer
David W. Taylor Naval Ship R&D Center

APPROVED FOR PUBLIC RELEASE; DISTRIBUTION UNLIMITED.

SHIP MATERIALS ENGINEERING DEPARTMENT

June 1979 DTNSRDC-79/063

MAJOR DTNSRDC ORGANIZATIONAL COMPONENTS



UNCLASSIFIED

SECURITY CLASSIFICATION OF THIS PAGE (When Data Entered)

REPORT DOCUMENTATION PAGE		READ INSTRUCTIONS BEFORE COMPLETING FORM
1. REPORT NUMBER DTNSRDC-79/063	2. GOVT ACCESSION NO.	3. RECIPIENT'S CATALOG NUMBER
4. TITLE (and Subtitle) FATIGUE FAILURE PREDICTION BY X-RAY DOUBLE CRYSTAL DIFFRACTOMETRY		5. TYPE OF REPORT & PERIOD COVERED RESEARCH AND DEVELOPMENT
		6. PERFORMING ORG. REPORT NUMBER
7. AUTHOR(s) R.N. Pangborn and S. Weissmann (Department of Mechanics, College of Engineering, Rutgers University) and I.R. Kramer		8. CONTRACT OR GRANT NUMBER(s)
9. PERFORMING ORGANIZATION NAME AND ADDRESS David W. Taylor Naval Ship R&D Center Annapolis, Maryland 21402		10. PROGRAM ELEMENT, PROJECT, TASK AREA & WORK UNIT NUMBERS (see reverse side)
11. CONTROLLING OFFICE NAME AND ADDRESS David W. Taylor Naval Ship R&D Center Annapolis, Maryland 21402		12. REPORT DATE June 1979
		13. NUMBER OF PAGES 15
14. MONITORING AGENCY NAME & ADDRESS(if different from Controlling Office) 24p.		15. SECURITY CLASS. (of this report) UNCLASSIFIED
		15a. DECLASSIFICATION/DOWNGRADING SCHEDULE

16. DISTRIBUTION STATEMENT (of this Report)

APPROVED FOR PUBLIC RELEASE: DISTRIBUTION UNLIMITED.

ZR00001

17. DISTRIBUTION STATEMENT (of the abstract entered in Block 20, if different from Report)

ZR0220101

18. SUPPLEMENTARY NOTES

19. KEY WORDS (Continue on reverse side if necessary and identify by block number)

Fatigue
X-ray
Aluminum alloy

20. ABSTRACT (Continue on reverse side if necessary and identify by block number)

X-ray double crystal diffractometry and reflection topography were employed to examine the deformation response of single crystals and Al 2024 alloy specimens. A propensity for work hardening in the surface layers compared to the bulk material was demonstrated for both tensile-deformed and fatigue-cycled metals. The cyclically induced defect distribution with depth from the specimen surface was also investigated as a function of the fraction of fatigue life. This led to a capability for predicting the fatigue life and failure by a nondestructive x-ray diffraction method.

DD FORM 1 JAN 73 1473 EDITION OF 1 NOV 65 IS OBSOLETE
S/N 0102-LF-014-6601

UNCLASSIFIED

SECURITY CLASSIFICATION OF THIS PAGE (When Data Entered)

387682

UNCLASSIFIED

SECURITY CLASSIFICATION OF THIS PAGE (When Data Entered)

(Block 10)

Program Element - 61152N
Project - ZR00001
Task Area - ZR0220101
Work Unit - 2802-004

UNCLASSIFIED

SECURITY CLASSIFICATION OF THIS PAGE(When Data Entered)

TABLE OF CONTENTS

LIST OF FIGURES

LIST OF ABBREVIATIONS AND SYMBOLS

Al	Aluminum
Cu	Copper
(hkl)	X-ray reflection planes
Mo	Molybdenum
MPa	Megapascal
N	Number of fatigue cycles
N_f	Number of fatigue cycles to failure
X	Distance
β_m	Width of the rocking curve at half the peak intensity maximum
ρ	Excess dislocation density
μm	Micrometers

ABSTRACT

X-ray double crystal diffractometry and reflection topography were employed to examine the deformation response of single crystals and Al 2024 alloy specimens. A propensity for work hardening in the surface layers compared to the bulk material was demonstrated for both tensile-deformed and fatigue-cycled metals. The cyclically induced defect distribution with depth from the specimen surface was also investigated as a function of the fraction of fatigue life. This led to a capability for predicting the fatigue life and failure by a nondestructive x-ray diffraction method.

ADMINISTRATIVE INFORMATION

This investigation is part of an in-house research program at the David W. Taylor Naval Ship Research and Development Center. It was conducted under task ZR0220101, work unit 2802-004.

INTRODUCTION

The ability to predict the failure of materials subjected to repeated or alternating load is of great technological importance, both for mechanical and structural applications. Because x-ray diffraction methods are nondestructive, they have been used extensively in the past to elucidate structural changes. In these previous investigations, however, the change in the patterns, such as asterism or line broadening, occurred either at the very early stages of cycling, remaining invariant thereafter,[1,2*] or were restricted to that period which follows the initiation of catastrophic macroscopic failure.[3,4] Consequently, the attempts to forecast ultimate failure at early stages of the life were unsuccessful.

We recently applied a special x-ray method which enabled us to predict the fatigue life and failure of Al 2024 with considerable reliability.[**] The nondestructive technique combines diffraction analysis using a double crystal diffractomer, with x-ray reflection (Berg-Barrett)

---

*A list of references is given on page 10.

**A list of symbols and abbreviations is given on page v.

topography to provide a visualization of the surface defect structure.[5,6] For polycrystalline specimens, the individual grains are considered to function independently as the second crystal of a double crystal diffractometer. Rocking curves for the reflecting grains are obtained by irridating the specimen with crystal monochromated radiation, and by recording the spot reflections on a cyclindrical film along the Debye-Scherrer arcs. By incremental specimen rotation, combined with film shifts to separate the spot contributions, arrays of spots are obtained. The intensity dependence of each spot array represents the x-ray rocking curve for the grain, and gives a measure of its dislocation density. The width of the rocking curve at half the peak intensity maximum is denoted by the measured halfwidth β_m. The halfwidth of grains for various (hkl) reflections can be averaged to give a representative halfwidth $\bar{\beta}_m$ for the grain population. Corrected for the halfwidth β_o of the virgin specimen, $\bar{\beta}$ is obtained by the relation $\bar{\beta} = (\bar{\beta}_m^{\,2} - \beta_o^{\,2})^{1/2}$. The excess dislocation density ρ for the specimen is then calculated from $\bar{\beta}$ according to the relation[7]

$$\bar{\rho} = \bar{\beta}^2/9b^2 \ . \qquad (1)$$

LATTICE DEFECTS IN SINGLE CRYSTALS INDUCED BY MONOTONIC DEFORMATION

It has been reported previously[8-11] that single crystals and commercial alloys show a propensity for preferential work hardening of the surface layer, when deformed by uniaxial tension or fatigue cycling. It was proposed that when the work hardening reached a critical value, which was independent of stress amplitude, cyclic history and environment, fracture was initiated if the accumulation of excess dislocations produced local stresses in excess of the fracture strength.

Our prediction of fatigue failure on the basis of x-ray rocking curves was developed from a systematic study of the excess dislocation distribution from the surface to bulk of tensile deformed single crystals. Three model materials were chosen: silicon, a low stacking fault energy, brittle material, which is ductile when deformed at elevated temperature, and for which the dislocations are virtually immobile when cooled to

ambient temperature; aluminum, a high stacking fault energy material, with microstructure typical of ductile fcc metals; and gold, which exhibits little propensity for surface oxide layer formation. The distribution of excess dislocations in depth was determined by rocking curve measurements after discrete surface layer removals by chemical dissolution or electropolishing.

As shown in Figure 1, a decreasing gradient of excess dislocation density from surface to bulk was obtained for all three species. It will be noted that there was a decline by factors of 2.7 to 6.2, from a high surface dislocation density $\bar{\rho}_s$ to a constant value $\bar{\rho}_i$ at about 150 μm into the interior. Although tested to only 3% plastic strain, the gold revealed a similar gradient. This result indicates that oxide film formation is not a necessary prerequisite for the decrease of excess dislocations from the surface to the interior. A relation proposed to describe the dislocation gradient in ground alkali halide crystals[12] can be employed in this study to convert the obtained curves of Figure 1 into the linear form

$$\ln\{(\bar{\rho}_x/\bar{\rho}_i) - 1\} = \ln\{(\bar{\rho}_s/\bar{\rho}_i) - 1\} - kx \qquad (2)$$

where $\bar{\rho}_x$ is the excess dislocation density at the depth distance x. The linear conversion yielded a consistent material-independent slip k of about 2.85 for all three materials.

LATTICE DEFECTS IN FATIGUE-CYCLED Al SINGLE CRYSTALS

Aluminum crystals were subjected to axial tension-compression fatigue, and the defect structure induced in the high cycle range (±1.03 MPa, 2×10^5 cycles) was studied in depth by rocking curve measurements. As shown in Figure 2, a decreasing gradient from the surface into depth, typical of monotonic deformation, was also produced by the cycling.

The depth profile exhibited, however, an additional salient feature. A minimum in the curve was obtained at a depth distance of 100 μm from the surface. Further in depth, the halfwidth value increased again and attained a plateau level at about 250 μm from the surface and extending into the bulk. It can be noted that the halfwidth values measured for the

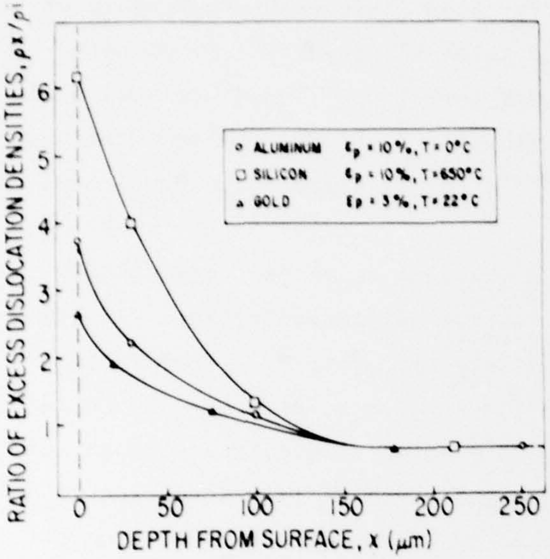


Figure 1 - Distribution of Excess Dislocations with Depth from the Crystal Surface, Expressed as the Ratio of the Density $\bar{\rho}_x$ at Each Depth to that of the Bulk $\bar{\rho}_i$. Tensile Axis and Surface Orientations:
Al,<100> and (100);
Si,<110> and (112);
Au,<123> and (311)

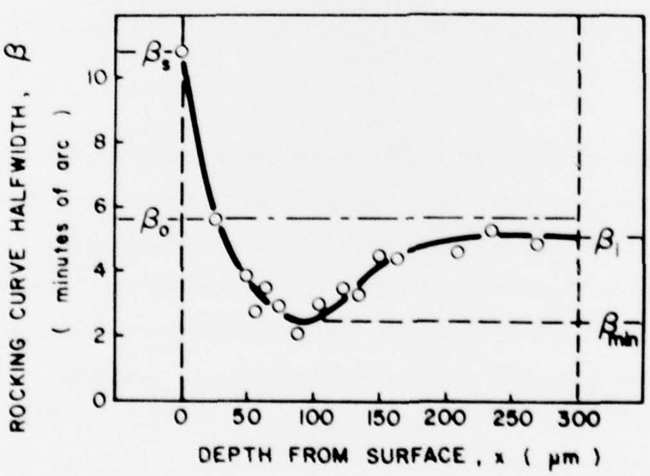


Figure 2 - Profile of the Measured Half-width Dependence on Depth from the Surface for a Fatigued Al Single Crystal; 2×10^5 Cycles at ± 1.03 MPa, <100>

subsurface layer extending from 50 to 150 μm in depth were lower than that of the virgin crystal, indicating that a cyclic recovery process occurred in this region. Furthermore, it may be seen that in the bulk no change from the intrinsic halfwidth was produced by the low amplitude cycling.

DEPTH DISTRIBUTION OF LATTICE DEFECTS IN FATIGUE-CYCLED Al ALLOY

Al 2024 alloys were cycled in the push-pull mode at an amplitude of ±200 MPa corresponding to the proportional limit in static tension. The fatigue behavior of the specimens is given in Table 1. As shown in Figure 3, Part A, the measured rocking curve halfwidths, and thus the derived excess dislocation densities, increased throughout the specimen with increased cycling. Analogous to the deformation characteristics of monotonically and cyclically deformed single crystals, the depth profile analysis of Part A in the figure revealed higher excess dislocation densities in the surface layer. After 0.15% of the fatigue life, a decreasing gradient to a constant bulk level was obtained, similar to that observed for simple tension (the first half cycle of push-pull fatigue). For 5% of the fatigue life, an increase to a bulk plateau level was again evident, as found previously for single crystal fatigue.

TABLE 1 - FATIGUE LIFE OF THE 2024 Al SPECIMENS; R = -1

Stress (MPa)	Cycles to Failure
200	21,100
150	191,000
100	6,740,000

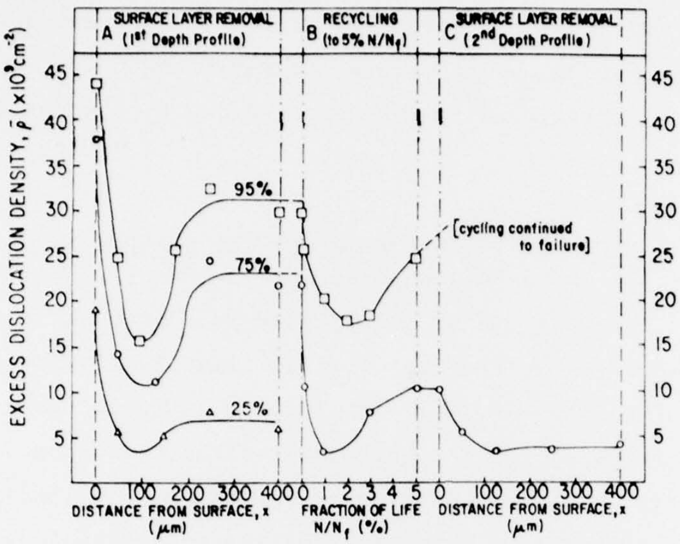


Figure 3 - Composite Diagram for Al 2024 Specimens Given Prior Fatigue to 75 and 95% of Their Life at ±200 MPa, Followed by Surface Removal (Part A) and Recycling (Part B), and Either Continued Recycling or X-ray Analysis in Depth (Part C)

Figure 4 shows data obtained by analysis of the surface layer alone, using shallow-penetrating copper radiation, after cycling to various fractions of the fatigue life. The results are plotted for specimens with two average grain sizes, differing by about 25% in grain diameter, and for tests carried out at two different stress amplitudes. The change in excess dislocation density during the life could be described by a three-stage sequence for all the tests. Rapid increases in the defect concentration occurred early (Stage I) and late (Stage III) in the life. A markedly decreased slope was obtained for the long duration of the second stage, comprising the period from 15 to 95% of the fatigue life. All the data fell within the experimental error band, even for the alternate grain size stock if a uniform shift factor related to the ratio of grain diameters was applied.

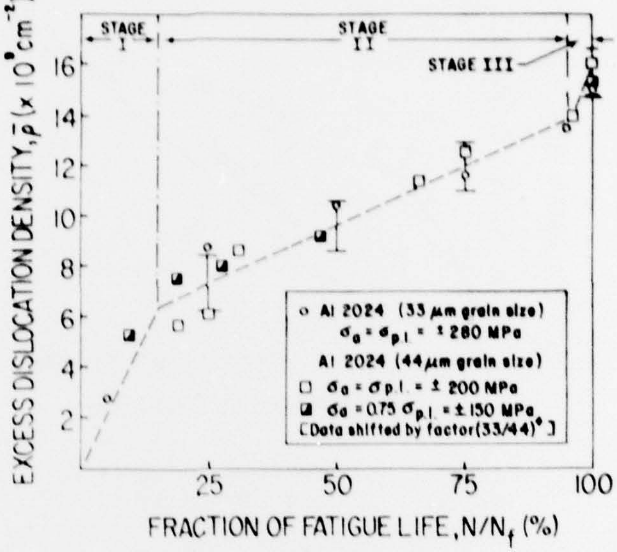


Figure 4 - Three-Stage Dependence of the Excess Dislocation Density in the Surface Layer During the Fatigue Life for Two Al 2024 Batches, Tested at Various Amplitudes

By carrying out halfwidth measurements as a function of the life fraction, utilizing deeply-penetrating molybdenum radiation, grains at depths up to 350 μm could be analyzed. Figure 5 shows a comparison of the excess dislocation densities obtained by analysis with Cu and Mo radiations, which differ in maximum penetration by an order of magnitude. Unlike the rapid saturation of the dislocation density for surface grains, as sampled by Cu radiation, analysis with Mo radiation provided a much steeper and nearly linear increase. At failure, the terminal dislocation density was identical to the critical excess dislocation density $\bar{\rho}^*$ measured using Cu radiation. This steeply inclined dependence of $\bar{\rho}$ with fatigue life permits a much more accurate prediction of the final failure. Referring to Figure 3, Part A, it may be seen that this result derives from the gradual increase of the plateau value of $\bar{\rho}$ in the bulk as a function of fatigue life, as compared to the rapid saturation at the surface layer. Only the Mo radiation was able to disclose the gradual increase of $\bar{\rho}$ in the interior.

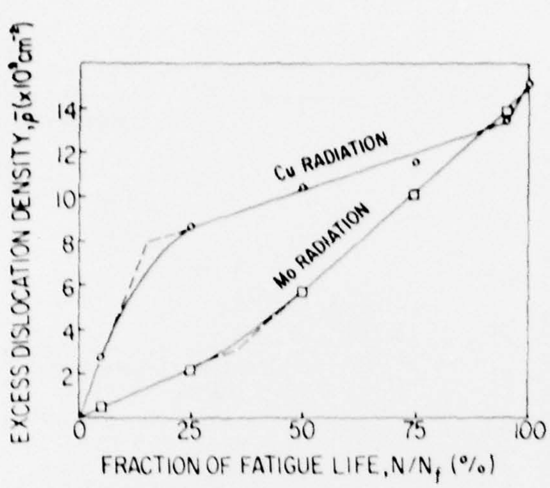


Figure 5 - Comparison of Excess Dislocation Densities Measured During the Fatigue Life by Analysis with Copper and Molybdenum Radiations. Maximum Penetration Distances: CuKα, 37 μm; Mokα, 350 μm

To test the stability of the defect structure in the grains of the bulk induced by cycling, a further experimental program was carried out. As shown in the composite diagram of Figure 3, the surface layers of previously cycled specimens were removed by electropolishing to a depth of 400 μm. The specimens given prior cycling to 75 and 95% of their life, followed by the surface removal, were then recycled at the same amplitude. The excess dislocation density declined rapidly during the initial recycling, as revealed in Part B of the figure. This remarkable experimental observation has to be attributed to the instability of the bulk structure on initial cycling when the work hardened surface is absent. After reaching a minimum at 1 to 2% of the life, the density increased again, since a new, work hardened surface layer was again being formed.

When the recycling process was interrupted after 5% to obtain a second depth profile, as exhibited in Figure 3, Part C, the analysis of the specimen given 75% prior cycling revealed an excess dislocation density over the entire cross section nearly equal to the value of the undeformed, virgin specimen. The instability of the bulk dislocation

structure when cycled in the absence of the work hardened surface layer explains the extension of fatigue life afforded by judicious, periodic surface layer removal.[13,14] Uninterrupted recycling of a specimen previously fatigued to 95% of its life and polished to 400 μm in depth, produced a 75% net increase in the normal fatigue life.

It is therefore apparent that the accumulation of prefatigue failure damage is dependent on a mutual interplay between the defect structures in the surface layer and bulk material. This study has also shown that the fatigue life can be characterized by the nearly linear dependence of excess dislocation density as measured by rocking curve measurements using penetrating radiation. Thus, by preestablishing the critical excess dislocation density ρ^* and comparing it to the ρ value measured at any stage in the fatigue process, the remaining fraction of fatigue life can be determined.

FUTURE WORK

In the continuation of this project it is intended to test the capability of this x-ray method to predict cumulative fatigue damage. The future investigation will also include a determination of the excess dislocation-depth profile and critical dislocation density for specimens cycled in bending ($R = -1$), tension-tension ($R - 0$), and under constant strain amplitude conditions. The applicability of the line profile method for measuring the excess dislocation will be investigated.

REFERENCES

1. Bennett, J.A., "A Study of Fatigue in Metals by Means of X-ray Strain Measurement," J. Res. NBS, Vol. 46, p. 457 (1951).

2. Taira, S. and K. Hayashi, "X-ray Investigation on Fatigue Fracture of Notched Steel Specimen," Bull. JSME, Vol. 9, p. 627 (1966).

3. Gough, M.J. and W.A. Wood, "A New Attack Upon the Problem of Fatigue of Metals, Using X-ray Methods of Precision," Proc. Roy. Soc. (London), Vol. A154, p. 510 (1936).

4. Spencer, R.G. and J.W. Marshall, "An X-ray Study of the Changes that Occur in Aluminum During the Process of Fatiguing," J. Appl. Phys., Vol. 12, p. 191 (1941).

5. Intrater, J. and S. Weissmann, "An X-ray Diffraction Method for the Study of Substructure of Crystals," Acta Cryst., Vol. 7, p. 729 (1954).

6. Weissmann, S. and D.L. Evans, "An X-ray Study of the Substructure of Fine-Grained Aluminum," Acta Cryst., Vol. 7, p. 733 (1954).

7. Hirsch, P.B., "Mosaic Structure," Prog. Met. Phys., Vol. 6, p. 283 (1956).

8. Kramer, I.R. and L.J. Demer, "Effect of Surface Removal on the Plastic Behavior of Aluminum Single Crystals," TMS-AIME, Vol. 221, p. 780 (1961).

9. Kramer, I.R., "Work Hardening and Creep of Aluminum and Copper During Alternate Stressing," Trans. ASM. Vol. 62, p. 521 (1969).

10. Kramer, I.R., "A Mechanism of Fatigue Failure," Met. Trans., Vol. 5, p. 1735 (1974).

11. Kramer, I.R., "Influence of the Surface Layer on the Plastic-Flow Deformation of Aluminum Single Crystals," TMS-AIME, Vol. 233, p. 1462 (1965).

12. Belykh, G.I., G.M. Pyatigorskii, and E.I. Raikhel, "Determination of the Parameter Representing the Dislocation Density from the Integrated X-ray Reflection Coefficient," Sov. Phys. Sol. St., Vol. 18, p. 161 (1976).

13. Thompson, N., N. Wadsworth, and N. Louat, "The Origin of Fatigue Fracture in Copper," Phil. Mag., Vol. 1, p. 113 (1956).

14. Kramer, I.R. and A. Kumar, "Relaxation and Cyclic Hardening of the Surface Layer of Copper," Met. Trans., Vol. 3, p. 1223 (1972).

INITIAL DISTRIBUTION

Copies	
3	CNR 1 Code 465 1 Code 471
3	NRL 1 Code 6311 1 Code 6380 1 Code 6385
2	NAVAIR (AIR 320)
1	Naval Air Development Center Naval Materials Center Warminster, PA 18974 Attn: Dr. F.S. Williams (Code 606)
5	NAVSEA 1 SEA 035 1 SEA 05D 1 SEA 323 2 SEA 99612
12	DDC
1	Air Force Materials Laboratory Wright-Patterson AF Base Dayton, OH 45433 (Attn: Dr. H. Burte)

CENTER DISTRIBUTION

Copies	Code	
2	012	
2	17	
1	28	
1	280	
15	2802	
2	281	
6	282	
10	5211.1	Rept Dist
1	522.1	Library (C)
1	522.2	Library (A)
2	5231	Off Services